

material for wheels.

The problematic described above is not new. Therefore, for many years there has been the wish for an Al alloy, which combines the high strength properties of the alloys AA 2014 or AA 2214 at ambient temperature and the thermal stability of the alloys AA 2618 or 2618 A.

The invention therefore addresses the problem of providing such an alloy, a semifinished product produced of such an alloy with high static and dynamic bearing capacity, high thermal stability, high fracture toughness and high creep resistance as well as a method for the production of such semifinished products.

This problem is solved according to the invention thereby that the alloy has the following composition:

0.3 – 0.7 wt. % silicon (Si)
maximally 0.15 wt. % iron (Fe)
3.5 – 4.5 wt. % copper (Cu)
0.1 – 0.5 wt. % manganese (Mn)
0.3 – 0.8 wt. % magnesium (Mg)
0.05 – 0.15 wt. % titanium (Ti)
0.1 – 0.25 wt. % zirconium (Zr)
0.3 – 0.7 wt. % silver (Ag)
maximally 0.05 wt. % other, individually
maximally 0.15 wt. % other, total
remaining wt. % aluminum (Al).

Compared to the prior known alloys AA 2014 and AA 2214, the claimed alloy has higher static and dynamic thermal stability and improved creep resistance with simultaneously very good mechanical fracturing properties. These are attained in particular at a copper-

magnesium ratio between 5 and 9.5, in particular at a ratio between 6.3 and 9.3. The copper content is preferably between 3.8 and 4.2 wt. % and the magnesium content between 0.45 and 0.6 wt. %. The copper content is markedly below the maximum solubility for copper in the presence of the claimed magnesium content. As a consequence, the fraction of insoluble copper-containing phases is very low also taking into consideration the remaining alloy and companion elements. Thereby an improvement is obtained with respect to the dynamic properties and the fracture toughness of the semifinished products manufactured from such an alloy.

In contrast to the known AA alloys 2014 and 2219, a portion of the claimed alloy is silver with contents between 0.3 and 0.7 wt. %, preferably 0.45 and 0.6 wt. %. In the interaction with silicon (0.3 – 0.7 wt. %, preferably 0.4 – 0.6 wt. %) the hardening takes place via the same mechanisms as in silver-free Al/Cu/Mg alloys. However, it has been found that with lower silicon contents, the course of precipitation is different due to the addition of silver. While the semifinished products manufactured from such an alloy have good high-temperature stability and creep resistances under cooler conditions, however, they do not meet the desired requirements. Only silicon contents above 0.3 wt. % suppress the otherwise typical change of the precipitation behaviour of Al/Cu/Mg/Ag alloys, such that unexpectedly higher strength values can be attained without having to give up the high-temperature stability and the creep resistance with the Cu and Mg contents according to the invention.

The manganese content of the claimed alloy is 0.1 to 0.5 wt. %, preferably 0.2 – 0.4 wt. %. In the case of alloys with higher manganese contents undesirable precipitation processes were found with long-term high-temperature stress, which led to a decrease of strength. For this reason the manganese content is limited to 0.4 wt. %. However, manganese is an alloy component fundamentally required for the control of the grain structure.

To balance the reducing effect of manganese with respect to the grain structure control, the

alloy contains zirconium between 0.10 – 0.25 wt. %, preferably 0.14 – 0.20 wt. %. The precipitating zirconium aluminides, as a rule, are developed even more finely dispersed than manganese aluminides. It was moreover found that the zirconium aluminides contribute to the thermal stability of the alloy.

For grain sizing 0.05 – 0.15 wt. %, preferably 0.10 – 0.15 wt. % of titanium is added. The titanium is usefully added in the form of an Al/5Ti/1B prealloy, whereby boron is automatically included in the alloy. Therefrom finely dispersed, insoluble titanium diborides are formed. These contribute to the thermal stability of the alloy.

The alloy can comprise maximally 0.15 % iron, preferably 0.10%, as an unavoidable contamination.

In the following, test results will be described with reference to the attached figures. These depict:

Fig. 1 a diagram representing the 0.2% yield strength and the tensile strength of the alloy according to the invention in state T6 in comparison to prior known alloys, as a function of the test temperature,

Fig. 2 a diagram representing the long-time stress to rupture strength of the alloy according to the invention in state T6 in comparison to known alloys,

Fig. 3 a diagram representing the 0.2% yield strength and the tensile strength of airplane wheels manufactured from the alloy according to the invention in comparison to such manufactured from known alloys, and

Fig. 4 diagrams representing the fatigue strength of the alloy according to the invention in comparison to a known alloy in state T6 at ambient

temperature and at a temperature of 200° C.

Table 1 reproduced below shows the chemical composition of four alloys (B, C, D, E) according to the invention as well as the composition of the alloys AA 2214 and AA 2618 examined as a comparison (data in wt. % (n.d.: not determined))

Table 1

Alloy	Si	Fe	Cu	Mn	Mg	Ni	Zn	Ti	Ag	Zr	V
B	0.47	0.08	4.40	0.200	0.58	0.003	0.048	0.135	0.45	0.150	0.018
C	0.47	0.08	3.64	0.210	0.59	0.003	0.015	0.115	0.52	0.150	0.017
D	0.47	0.08	3.87	0.200	0.61	0.003	0.015	0.117	0.52	0.150	0.019
E	0.52	0.08	4.14	0.200	0.61	0.003	0.02	0.115	0.44	0.150	0.018
AA 2214	0.77	0.17	4.29	0.883	0.57	0.003	0.031	0.024	0.003	0.007	n.d.
AA 2618	0.22	1.1	2.58	0.020	1.53	1.007	0.043	0.059	0.003	0.002	n.d.

From these alloys semifinished products were manufactured following the method steps listed below:

- casting of an ingot from an alloy,
- homogenizing the cast ingot at a temperature, which is as close under the incipient melting temperature of the alloy as is possible, for a length of time adequate to attain maximally uniform distribution of the alloy elements in the cast structure,
- hot working of the homogenized ingot by forging at a block temperature of approximately 420°C,
- solution treatment of the semifinished product worked by forging at temperatures sufficiently high to bring the alloy elements necessary for the hardening into solution such that they are uniformly distributed in the structure, with the solution treatment taking place in a temperature range [sic] of 505°C over a time period of 3 hours,
- quenching of the solution-treated semifinished product in water at ambient temperature,
- cold working of the quenched semifinished products by cold upsetting by 1 to 2%, and
- artificial ageing of the quenched semifinished product at a temperature of 170°C over a

time period of 20 to 25 hours.

The open-die forged pieces produced in this manner were subsequently tested for their properties in the artificially aged state T6.

Table 2

Strength values at ambient temperature Fracture toughness at ambient temp.

Alloy	Sample direction	R _{p02} (MPa)	R _m (MPa)	A ₅ (%)	Sample direction	K _{IC} (MPa√m)
C	L	448	485	11.2	T - L	31.3
	LT	427	471	7.2	S - L	29.5
	ST	417	479	6.3	S - T	32.2
D	L	456	495	10.7	T - L	28.3
	LT	434	478	8.0	S - L	29.1
	ST	429	484	5.5	S - T	29.6
E	L	454	494	9.9	T - L	26.1
	LT	446	493	6.4	S - L	25.5
	ST	438	494	4.9	S - T	26.9
AA 2214	L	444	489	9.7	T - L	24.2
	LT	439	483	6.4	S - L	25.9
	ST	429	480	5.8	S - T	27.3
AA 2219	L	286	408	16.7	T - L	31.1
	LT	288	403	8.4	S - L	34.4
	ST	366	455	5.0	S - T	32.3
AA 2618	L	389	443	5.1	T - L	19.2
	LT	383	437	4.7	S - L	16.7
	ST	376	427	4.1	S - T	19.3

Table 3

<i>Alloy</i>		<i>E</i>			<i>AA 2214</i>			<i>AA 2618</i>		
R_{test} (°C)	T_{hold} (h)	$R_{p0.2}$ (Mpa)	R_m (Mpa)	A_5 (%)	$R_{p0.2}$ (Mpa)	R_m (Mpa)	A_5 (%)	$R_{p0.2}$ (Mpa)	R_m (Mpa)	A_5 (%)
20	1	454	494	9.9	444	489	9.6	380	434	6.5
50	1	453	493	12.6	443	485	9.8	382	433	6.1
100	1	449	474	13	425	458	11	374	423	6.5
150	1	404	417	14.3	403	424	13.6	366	404	7.6
170	1	403	416	16.3	382	400	13.6	382	389	9.6
200	1	355	372	18	348	368	13.8	340	359	12.2
220	1	340	351	18	324	344	14.2	301	332	12.4
250	1	268	282	19	250	268	16.1	282	300	14.7

Definitions sample directions:

L = longitudinal direction: parallel to the main form change direction

LT = long transverse direction: parallel to the width direction

ST = short transverse direction: parallel to the thickness direction

The improved strengths of the alloy according to the invention (for example alloy E) is clearly evident in Tables 2 and 3. For example, while the prior known alloy AA 2214 shows good strength values at ambient temperature, however at higher temperatures it does not. Moreover, the creep resistance as well as the fracture toughness are not only markedly better at ambient temperature, but especially also at higher temperatures, in the claimed alloy than in the prior known alloys. This comparison makes further clear that the tested prior known alloys have only good properties with respect to a single strength parameter. Not in a single case do these have good properties in all relevant strength values at ambient temperature as well as also at increased temperatures. Just as is the case with the fatigue properties, the creep resistance of this prior known alloy is not satisfactory. Very good properties over all tested strength parameters could only be determined in the case of the alloy according to the invention.

The associated representation of Figure 1 also makes graphically clear the better strength properties of the alloy (alloy E) according to the invention compared to the known alloys (AA 2214 as well as AA 2618). The results showed unexpectedly that the strength values

of alloy E are better even at temperatures below 100°C than those of the known alloy AA 2214, known for its especially high strength values in this temperature range.

Moreover, the creep resistance of the semifinished products was tested. Table 4 shown below provides the test results (LMP: Larson Miller parameter) in summary:

Table 4:

<i>Alloy</i>											
E				AA 2214				AA 2618			
T _{test} (°C)	σ _{test} (MPa)	t _{fracture} (h)	LMP (-)	T _{test} (°C)	σ _{test} (MPa)	t _{fracture} (h)	LMP (-)	T _{test} (°C)	σ _{test} (MPa)	t _{fracture} (h)	LMP (-)
180	185	2513	10.60	205	200	30	10.27	205	183	10	10.04
	167	4762	10.82		190	50	10.38		179	50	10.38
					181	100	10.52		175	100	10.52
					130	500	10.85		163	500	10.85
					100	800	10.95		159	1000	11.00

Plotted graphically, the markedly better long-time stress to rupture strength of the alloy in the T6 state in comparison to the known alloys AA 2214 and AA 2618 also in the T6 state is apparent. This is reproduced in the diagram of Figure 2 as time-compensated temperature representation. The especially good creep resistance of the alloy according to the invention could not be foreseen, such that this result is surprising.

Within the scope of testing the method steps for the production of these semifinished products, it was found that comparable material properties of the produced semifinished products can be attained if the step of hot working is carried out at a block temperature between 320°C to 460°C. The step of quenching of the solution treated semifinished product can take place in a temperature range between ambient temperature and 100°C (boiling) in water. It is also possible to utilize a water-glycol mixture for the quenching, the temperature of which, however, should not exceed 50°C. Instead of the previously described step of cold working through cold upsetting during forging, as a cold working

step also a drawing out by 1% to 5% can be carried out in the case of extruded or rolled products for the purpose of reducing the intrinsic stresses due to the quenching. The step of artificial ageing can be carried out over a time period of 5 to 35 hours, preferably between 10 and 25 hours in a temperature window between 170°C and 210 °C.

During further tests strand-cast ingots were produced as described above and airplane wheels manufactured by drop forging in the preforge die and finish forge die at a temperature of 410 to 430°C. These wheels were subsequently solution treated at 505°C, quenched in a mixture of water and glycol of ambient temperature and thermally age-hardened at 170°C for 20 hours. For comparison mass-produced airplane wheels of the alloy AA 2214 were used. At sites distributed over the circumference samples were removed from the wheels produced of the claimed alloy and of the conventional alloy, and tested for their tensile strength. The result is graphically shown in Figure 3. It can be clearly seen that the alloy E according to the invention yields better values compared to the known alloy AA 2214.

Fatigue tests in comparable samples of the two cited alloys also show that the wheels produced from the claimed alloy attain markedly better values than the wheels produced from the alloy AA 2214. This applies to the fatigue tests carried out at ambient temperature (cf. Figure 4a) as well as to the fatigue tests carried out at a test temperature of 200°C (cf. Figure 4b).

The description of the claimed invention surprisingly makes clear that these have not only high dynamic and static strength values, but that these have in particular also an especially good high-temperature stability, fracture toughness and creep resistance. Therefore this alloy is in particular suitable for the production of semifinished products, which must meet precisely these requirements, such as for example airplane wheels or compressors.